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Microstructure and wear studies of laser clad Al-Si/SiC_(p) composite coatings

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Abstract

Coatings of a composite material consisting of an Al-Si matrix reinforced with SiC particles were produced by laser cladding on UNS A03560 cast Al-alloy substrates from mixtures of powders of Al-12 wt.% Si alloy and SiC. The influence of the processing parameters on the microstructure and abrasive wear resistance of the coatings was studied. For an interaction time of 0.08 s and a power density of 330 MW/m², corresponding to a specific energy of 26 MJ/m², the interaction between SiC and liquid Al is limited and the reinforcement particles remain essentially undissolved. The coating's microstructure is formed of SiC particles dispersed in a matrix consisting of primary α -Al dendrites and interdendritic α -Al+Si eutectic. For interaction times of 0.3 and 0.45 s and a power density of 193 MW/m², corresponding to specific energies of 58 and 87 MJ/m², SiC reacts with molten Al and partially dissolves. The resulting microstructure consists of undissolved SiC particles, found mainly at the bottom of the clad tracks, where the maximum temperature reached during processing is lower, and Al₄SiC₄ and Si particles dispersed in a matrix of α -Al+Si eutectic. The coatings prepared with higher specific energy (58 MJ/m²) present a hardness of 250 V and an abrasive wear rate in three-body abrasion tests with SiC as abrasive of 1.7×10^{-4} mm³/m, while those produced with 26 MJ/m² present a hardness of 120 V and a wear rate of 0.43×10^{-4} mm³/m. These results show that Al₄SiC₄ and Si increase the hardness of the material by dispersion hardening but do not contribute to its abrasive wear resistance, because they are softer than the abrasive particles, and confirm that the parameters used to prepare Al-Si-SiC composite coatings by laser cladding must be selected so that only minimal reactions occur between SiC and molten Al. © 2007 Elsevier B.V. All rights reserved.

Keywords: Al-Si/SiC composite coatings; Al-Si-C formation; Laser cladding; Abrasive wear

1. Introduction

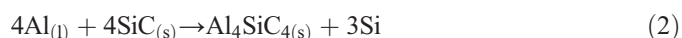
Due to their high strength, hardness, and wear resistance, aluminium-based metal matrix composites (Al-MMCs) find application in various industrial sectors, such as the automotive and aerospace industries [1]. The major drawbacks of these materials are their high cost and complex production methods compared to conventional alloys, but for many applications, like rapid tooling, the bulk stress levels are compatible with the use of a high-strength Al alloy, the required wear resistance being achieved by coating the component with a high wear resistance material such as a ceramic-reinforced Al-matrix composite. This can reduce material consumption and final cost.

SiC, B₄C, and TiC are excellent reinforcement materials for Al-MMCs, due to their high hardness and chemical and thermal stability [2–5], but carbides, in particular SiC, tend to react and

dissolve in molten Al, leading to the formation of Al₄C₃ and ternary Al-Si-C carbides during solidification [6]. Al₄C₃ has been widely reported in Al-SiC composites produced by melt-stir casting, in which the temperatures involved are relatively low (under 1200 K) [7,8]. Al₄C₃ and Al₄SiC₄ were also observed in Al-SiC coatings produced by laser cladding [9] and in welded joints in Al-SiC composites [10,11]. These compounds may result from chemical reactions between SiC and molten Al, which were studied by Viala et al. [6] over a wide temperature range (300 to 1900 K). The authors found that Al does not react with SiC up to its melting temperature (933 K), but molten Al reacts promptly with SiC, giving different products depending on temperature. Between 940 and 1620 K, Al₄C₃ is produced according to the reaction:



When the temperature exceeds 1670 K, the reaction product is the ternary carbide Al₄SiC₄, formed by the reaction:



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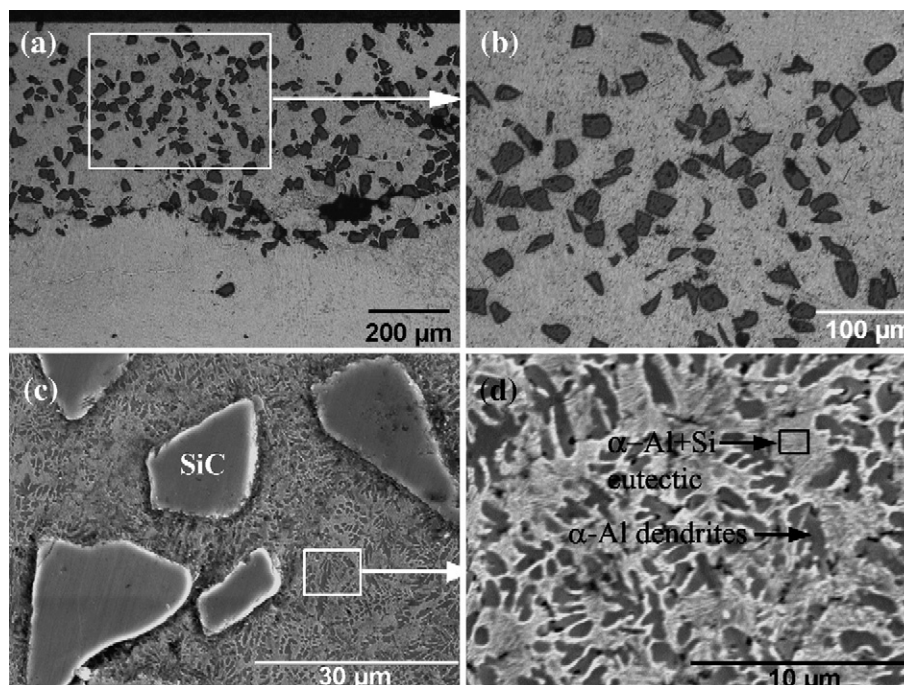


Fig. 1. Micrographs of coating prepared using a power density of 330 MW/m² and an interaction time of 0.08 s: (a) cross-sectional view of the coating; (b) and (c) SiC particles dispersed in matrix; (d) microstructure of the matrix.

At still higher temperatures (above 2200 K), the formation of Al₈SiC₇ in addition to Al₄SiC₄ has been reported [12]. The Si and C resulting from the partial dissolution of SiC and from these reactions also dissolve in liquid Al, increasing the saturation of the melt. As a result, Al₄C₃ and Al₄SiC₄ may also precipitate directly from the C and Si enriched Al-based melt formed [12].

The presence of Al₄Si₂C₅ carbide has been reported by Dahotre et al. [13] in pulsed laser welding of SiC/Al-alloy composites and by Liechti [14] that observed the presence of both Al₄SiC₄ and Al₂SiC₅ in Al-Si/SiC coatings produced by laser cladding. However, Al₄Si₂C₅ was not found in the detailed and extensive study carried out by Viala et al. [6] and Oden et al. [12] and controversy exists about the formation of this carbide.

Ureña et al. [11] studied the properties of aluminium and aluminium-silicon carbides and reported that Al₄C₃ is softer than Al₄SiC₄ (300 and 1200 HV, respectively) but, unlike Al₄SiC₄, it is brittle and tends to react with water, forming aluminium hydroxide. Accordingly, the presence of Al₄C₃ in the composite's microstructure results in poor mechanical properties and low long-term stability [8] and its formation must be avoided. Viala et al. [6] suggested that to achieve this goal, Si should be added to molten Al in order to displace reaction (1) to the left-hand side. The effectiveness of this method was demonstrated in the preparation of Al-SiC composites by the pressure infiltration technique [15], but it did not lead to the desired results in composite synthesis by laser particle injection, probably because in laser processing reactions do not occur in equilibrium conditions [16]. A similar behaviour should be expected for the reaction leading to the formation of Al₄SiC₄.

The laser-assisted preparation of Al-SiC composite coatings has been studied by several researchers [9,14,17–19]. Hu et al.

[17] prepared composite SiC-AA6061 alloy layers on substrates of the same alloy using laser powder injection and combinations of preplaced powder and blown powder laser cladding. The authors observed that the incorporation of SiC in coatings produced by laser powder injection was insufficient, in particular when the size of the reinforcement particles was large. The volume fraction of SiC incorporated into the microstructure was larger in coatings produced by preplaced powder laser cladding than in those produced by laser-assisted powder injection. Cladding with SiC-AA6061 alloy powder mixtures using the blown powder technique enabled large SiC particles to be incorporated into the composite, but these particles tended to crack and the coatings were porous. As a

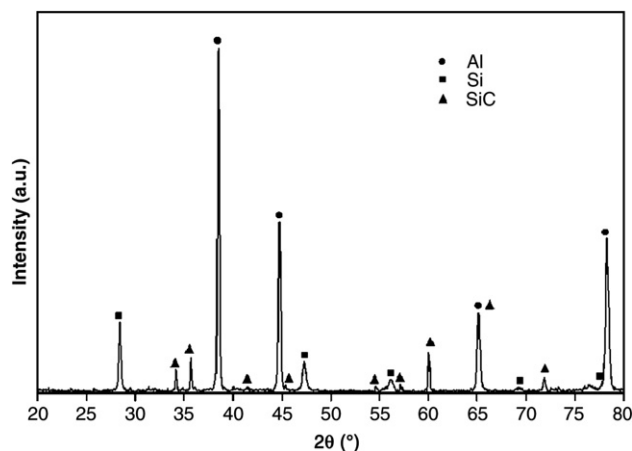


Fig. 2. XRD diffractogram of coating prepared using a power density of 330 MW/m² and an interaction time of 0.08 s.

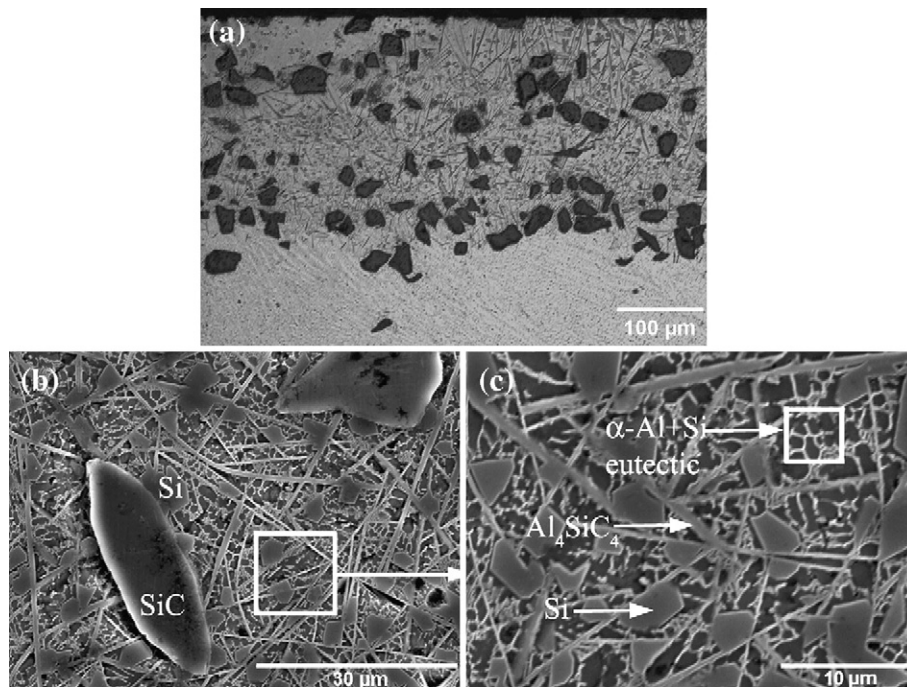


Fig. 3. Micrographs of coating prepared using a power density of 193 MW/m² and interaction time of 0.3 s: (a) cross-sectional view of the coating; (b) and (c) magnification showing plate-like particles, faceted particles and eutectic.

result of these defects, their sliding wear resistance was lower than that of coatings containing smaller reinforcement particles prepared by the preplaced powder technique. Al₄SiC₄, Al₄C₃ and Si were identified in the coating's microstructure. In a later paper [9], the same authors studied the influence of processing parameters on the coating's microstructure and properties. They observed that the aluminium carbide formed at low energy densities (100–200 MJ/m²) was Al₄C₃, while a high energy density (560 MJ/m²), leading to higher melt pool temperatures, resulted in the formation of Al₄SiC₄, in agreement with the conclusions of the Al–SiC equilibria study carried out by Viala et al. [6]. Hu et al. [9] carried out pin-on-disc sliding wear tests on SiC-AA6061 composite coatings produced by preplaced powder laser cladding with an average hardness of 380 HV_{0.1}, but the wear rate or the wear coefficient of the materials was not evaluated. The authors found that the weight loss of the composite coatings in tests performed at a load of 5 N was much lower than the weight loss of the substrate alloy. The SiC distribution in large area coatings produced by partially overlapping clad tracks was heterogeneous, the overlapping regions being depleted in SiC, but wear resistance did not significantly decrease as a result. No detailed studies of the sliding wear mechanisms or of the abrasive wear behaviour were performed. Selvan et al. [18] prepared similar coatings using the same technique and achieved surface hardnesses between 350 and 680 HV. Microscopic examination of the specimens showed that SiC particles were completely dissolved in the melt for the experimental conditions used, thus explaining to some extent the high hardness values. Vreeling et al. [19] studied in detail the preparation of SiC–Al surface composites by laser powder injection and showed that the injection depth of SiC particles depends more on the melt pool temperature than

on the initial velocity of the particles. This is because a continuous oxide film forms at the surface of the melt pool at low temperatures, which acts as a barrier against the penetration of SiC particles. At higher temperatures the oxide film partially dissolves, enabling particles to penetrate into the liquid. Consistently with this mechanism, the homogeneity of the SiC distribution could be improved by preheating the substrate to temperatures above 300 °C. However, due to the temperature gradient in the transverse section of the melt pool, sufficient particle penetration was only achieved in the central region of the laser tracks, the outer regions remaining essentially SiC-free. This is a major limitation when coating large areas by single track overlapping. The carbide present in the microstructure was Al₄C₃ [16].

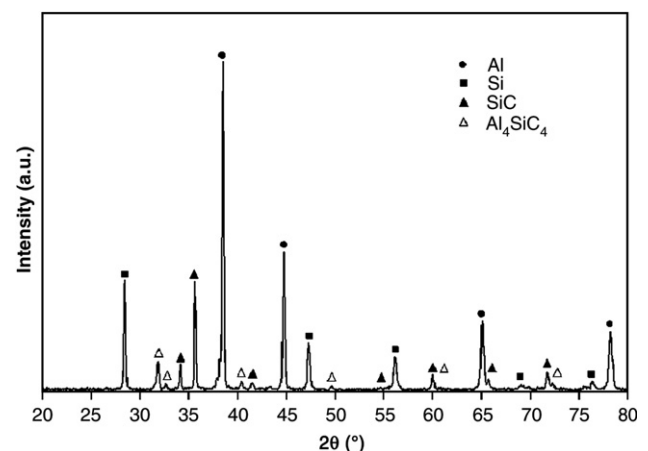


Fig. 4. XRD diffractogram of coating prepared using a power density of 193 MW/m² and an interaction time of 0.3 s.

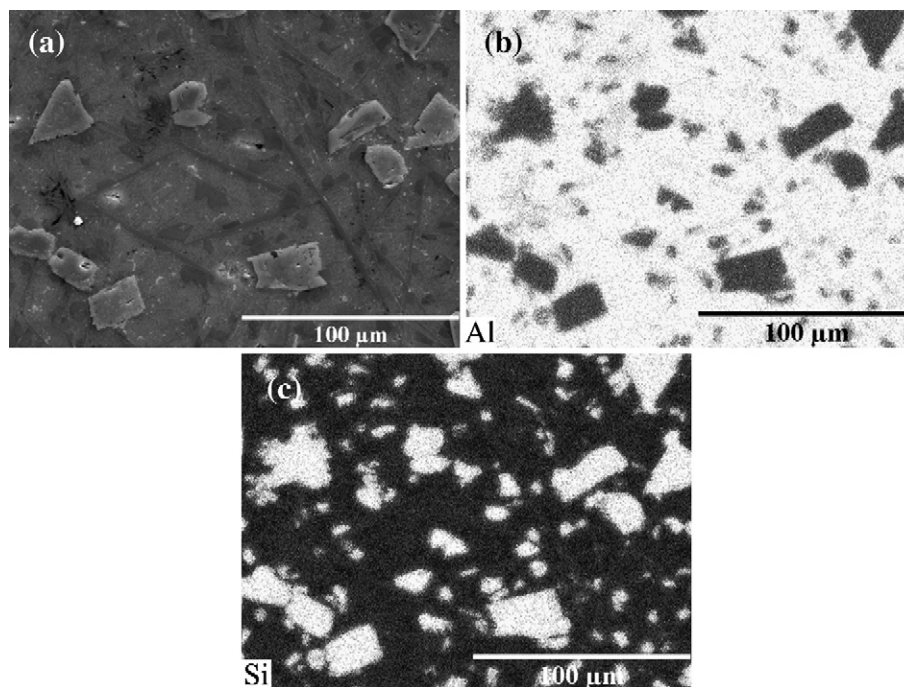


Fig. 5. EDS element distribution maps of coating prepared using a power density of 193 MW/m^2 and an interaction time of 0.3 s.

The results discussed so far suggest that laser powder injection lacks the robustness required for the industrial production of Al-SiC composite coatings. Laser cladding is at a clear advantage here because it presents a wider processing window [20]. As a result, processing parameters may be chosen so that the maximum temperature attained and the contact time between the reinforcement material and the melt are limited, thus restricting any reactions between them. Furthermore, the inherent flexibility of the process means the microstructure and properties of the coating can be tailored to specific requirements.

Liechti [14] studied Al-Si-SiC composite coatings deposited on Al-alloy substrates by the blown powder laser cladding technique. Al-12Si and Al-40Si alloy powders mixed with 17 and 35 vol.% SiC were used as starting materials. For the processing conditions used, microstructures containing Al_4SiC_4 , $\text{Al}_4\text{Si}_2\text{C}_5$ and Si along with SiC resulted. A hardness of 220 HV_{30} and a Young's modulus of 148 GPa were reported for coatings with Al40Si-35 vol.% SiC nominal composition. The coatings were free of major defects, but their wear resistance was not investigated.

In the present work the influence of processing parameters on the extent of the reactions between SiC and molten aluminium and hence on the microstructure and properties of Al-12Si/SiC coatings prepared by laser cladding was investigated, with emphasis on their abrasive wear resistance. The prevailing abrasive wear mechanisms for the different microstructures achieved were identified.

2. Experimental details

The laser cladding experiments were carried out using a 2 kW continuous wave Nd:YAG laser ($\lambda = 1.064 \text{ μm}$). Plates of

a commercial Al-7 wt.% Si cast alloy (UNS A03560) were used as substrates. Prior to deposition, the substrates were sand-blasted and cleaned in alcohol in an ultrasonic bath. The composite was synthesized in situ by independently feeding powders of Al-12 wt.% Si alloy (to form the matrix, 75 μm) and SiC (the reinforcement material, 45 μm) at mass flow rates in the ranges $0.04\text{--}0.09 \text{ g/s}$ and $0.02\text{--}0.04 \text{ g/s}$, respectively. Before being fed the powders were premixed and separated from part of the carrying gas in a cyclone. Powder feeding was carried out using a lateral nozzle kept at an angle of 45° to the horizontal. The velocity of the injected particles was varied in the range 0.5 to 5 m/s and controlled by opening or closing the gas outlet of the cyclone. The laser was incident at an angle of 80° to the substrate to prevent damage to the optical fibre and the laser cavity by the reflected beam. The powder-feeding nozzle allows for a co-axial flow of argon, not only to protect the melt pool from the atmosphere, but also to limit spreading of the powder jet. During the deposition experiments the powders in the hoppers were kept at a constant temperature of 40°C in order to reduce moisture and improve flowability. Deposition was carried out using a laser beam power of 1.75 kW , scanning speeds varying from 7.5 to 33.3 mm/s , a proportion of SiC in the powder mixture of 30 wt.% and with the distance between the sample surface and the focal plane of the converging optical system varying from 8 to 13 mm . The sample surface was located below the focal plane of the focusing optics. Using these parameters power densities and interaction times between 193 and 330 MW/m^2 and 0.08 and 0.45 s , respectively, were achieved. Coatings were produced by lateral overlapping of consecutive tracks by 30% of their width.

Transverse sections of the coatings were cut, metallographically polished and etched in Keller's reagent. Optical microscopy, scanning electron microscopy (SEM), energy

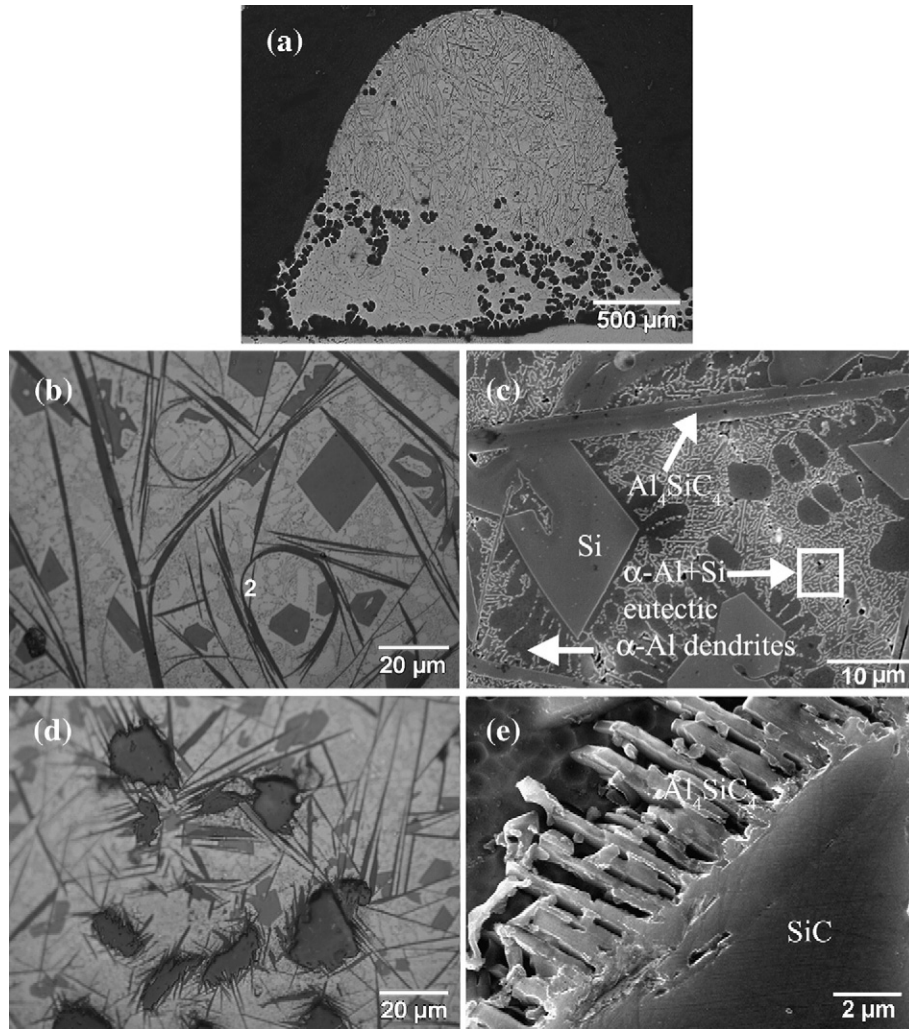


Fig. 6. Micrographs of clad prepared using a power density of 193 MW/m² and interaction time of 0.45 s; (a) cross-sectional view of the clad; (b) and (c) microstructure of the upper region of the clad; (d) and (e) microstructure of the lower region of the clad.

dispersive spectrometry (EDS) and X-ray diffraction were used for microstructural characterisation. The X-ray diffractograms were recorded in Bragg-Brentano configuration, using Cu α radiation. The volume fraction of SiC in the microstructure was determined by image analysis performed on optical micrographs. Specimens of the UNS A03560 alloy were also tested for comparison purposes. Hardness and abrasive wear tests were performed on the surface of the coatings, after metallographic polishing. The hardness measurements were performed at the coatings surface using a Vickers hardness tester and a load of 20 N. Microhardness tests were also performed in the cross-section of the coatings using loads of 20 N and 1 N. The microscale abrasive wear tests were performed using a ball cratering device [21] and a suspension of 35 wt.% SiC particles (4.25 μ m average diameter) in water as abrasive medium. The counterbody was a 19 mm diameter sphere of quenched and tempered AISI 440C tool steel (800 HV), which rotates against the surface of the sample to be tested. A constant sliding distance of 150 m and a normal load of 0.1 N were used. The volume of material removed (V) was determined from the

measurement of the wear crater diameter (d) using the equation [21]:

$$V = \frac{\pi d^4}{64R} \quad (3)$$

where R is the radius of the steel sphere. These values were used to calculate the material wear rate (volume of material removed per unit sliding distance). In order to investigate the wear mechanisms involved the worn surfaces were observed by scanning electron microscopy.

3. Results

3.1. Microstructure

The microstructure of the deposited material depends strongly on the laser processing parameters. A clad prepared with a power density of 330 MW/m² and a short interaction time of 0.08 s, corresponding to a specific energy of 26 MJ/m²,

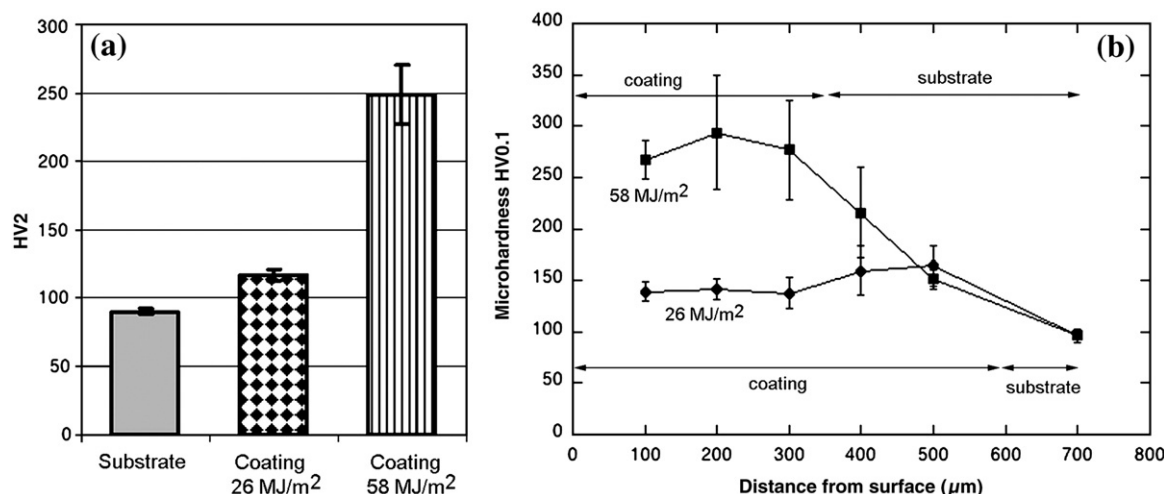


Fig. 7. (a) Surface hardness of coatings and substrate; (b) microhardness depth profile in the transverse cross-section of the coatings.

presents a microstructure essentially consisting of SiC particles dispersed in an aluminium alloy matrix (Fig. 1a). The matrix is formed of primary α -Al dendrites and α -Al+Si eutectic (Fig. 1d), as confirmed by X-ray diffraction (Fig. 2). The volume fraction of SiC is 28 ± 3 %.

Increasing the interaction time and the specific energy drastically modifies the microstructure. For example, the clad prepared with an interaction time of 0.3 s and a power density of 193 MW/m^2 (corresponding to a specific energy of 58 MJ/m^2) presents a lower volume fraction of SiC (18 ± 2 %) (Fig. 3a). The microstructure presents plate-like and faceted particles of two other phases dispersed in an α -Al+Si eutectic matrix (Fig. 3b and c).

These phases were identified on the basis of their morphology, X-ray diffraction and EDS analysis. The X-ray diffractograms presented in Fig. 4 show that the phases present in the structure are α -Al, Si, SiC and Al_4SiC_4 ternary carbide. These results, together with the EDS element distribution maps shown in Fig. 5, identify the plate-like particles as Al_4SiC_4 and the faceted equiaxed particles as Si. The Si particles are usually associated with SiC or to Al_4SiC_4 platelets, suggesting that Si nucleates at the interface between the carbide and liquid aluminium (Fig. 3b and c).

For the same laser power density and an interaction time of 0.45 s (resulting in a specific energy of 87 MJ/m^2), only a few SiC particles remain undissolved at the bottom of the clad (Fig. 6a). The microstructure in the upper region of the clad consists entirely of Al_4SiC_4 platelets and faceted Si particles, dispersed in a matrix of α -Al+Si eutectic (Fig. 6b and c). As in the previous case, Si nucleated on Al_4SiC_4 platelets, but in this specimen the Si equiaxed crystals are larger and surrounded by halos of α -Al with a dendritic morphology (Fig. 6c). The SiC particles at the bottom of the clad track are surrounded by a layer of reaction products, mainly Al_4SiC_4 (Fig. 6d and e). This layer is present only in regions where the maximum temperature attained was relatively low. Elsewhere the Al_4SiC_4 platelets do not show any relation to the SiC particles (Fig. 6b). In general, the clads prepared using these parameters did not show good

bonding to the substrate and no coatings were produced for mechanical characterisation using these parameters.

3.2. Hardness and abrasive wear resistance

Aiming to obtain average hardness values comparable with the wear testing results, the hardness of the coatings prepared with 26 and 58 MJ/m^2 were measured at the surface with a load of 20 N, leading to values of 120 and 250 HV respectively (Fig. 7a) in comparison to an average hardness of 90 HV for the substrate. These results were confirmed by measurements in the cross-section of the coatings with a load of 20 N, which lead to $140 \pm 20 \text{ HV}$ and $270 \pm 20 \text{ HV}$ for the coatings prepared with 26 and 58 MJ/m^2 respectively. The microhardness profiles in the transverse cross-section of the coatings, measured with a load of 1 N are presented in Fig. 7b. The results show that there is no significant hardness variation in the coatings with depth.

The abrasive wear rate is presented in Fig. 8. Both coatings present a lower wear rate than the substrate. Despite its higher hardness, the wear rate of the coating prepared with 58 MJ/m^2 is

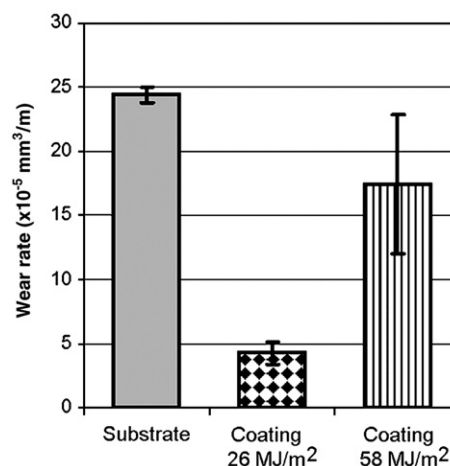


Fig. 8. Abrasive wear rate of the coatings.

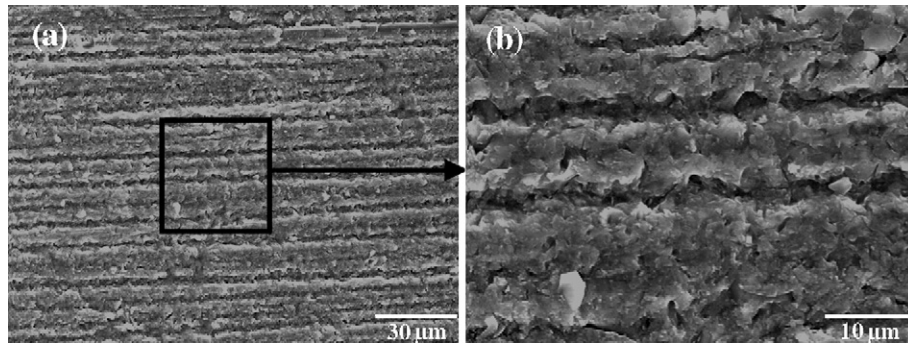


Fig. 9. SE micrographs of worn surfaces of the substrate Al-7Si alloy.

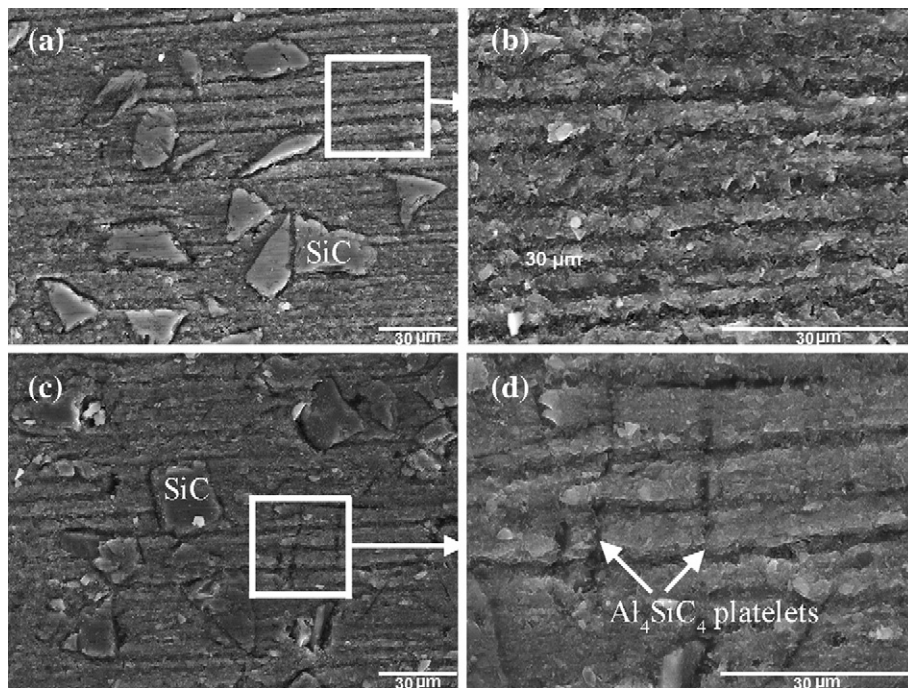


Fig. 10. SE micrographs of worn surfaces of the coatings prepared using specific energies of (a) and (b) 26 MJ/m² and (c) and (d) 58 MJ/m².

significantly higher than that of the coating prepared with 26 MJ/m².

The SEM micrographs of the worn surfaces of the substrate and the coatings are presented in Figs. 9 and 10, respectively. The worn surface of the Al-7Si alloy used as substrate presents continuous parallel grooves with a width similar to the average abrasive particle size (4.25 μm) (Fig. 9a). The surface presents evidence of intense plastic deformation and the grooves exhibit ridges along their edges (Fig. 9b). The worn surfaces of the composites (Fig. 10) also present abrasion grooves but these are discontinuous and confined to the matrix, because the hard SiC reinforcement particles interrupt the abrasive grooving action (Fig. 10a and c). The lateral ridges in the grooves are less conspicuous here than in the substrate, in particular in the coating prepared with higher specific energy. Also, the surface exhibits no traces of intense plastic deformation, as observed in the substrate. After the wear test the reinforcement particles remain adherent to the matrix, protruding from the surface. No fracture or detachment of these particles is observed. Micro-

graphs of Fig. 10c and d clearly show that Al₄SiC₄ platelets are scratched by the abrasive particles, offering no resistance to their grooving action.

4. Discussion

The microstructure of the coatings depends strongly on the processing parameters, particularly the interaction time between the laser radiation and the material and the radiation power density. When the interaction time is short (0.08 s, corresponding in the present case to a specific energy of 26 MJ/m²), no significant reaction between SiC and molten Al occurs because the contact between the two phases is brief and the melt pool temperature is relatively low. As a result, the composition of the melt remains essentially unchanged. During cooling this liquid, which consists of Al with 12 wt.% Si, solidifies as primary α-Al dendrites and α-Al+Si eutectic, instead of the fully eutectic microstructure predicted by the Al-Si equilibrium diagram for this composition. This non-equilibrium

microstructure has been previously observed in laser processed Al-Si eutectic alloys [22,23] and can be explained by the coupled zone theory. In a phase diagram the coupled zone represents the ranges of composition and temperature where the liquid solidifies by a eutectic reaction. In equilibrium conditions this occurs for a single invariant point in binary alloys, but in non-equilibrium conditions the eutectic reaction is kinetically favoured over the equilibrium pro-eutectic reactions in a range of compositions and temperatures represented by the coupled zone [24]. According to Fisher and Kurz [24], the coupled zone in the Al-Si system is shifted towards hypereutectic compositions because silicon has a faceted solid/liquid interface. As a result, the solidification of alloys with eutectic composition will start with the precipitation of α -Al dendrites for sufficiently high undercooling, as is the case in laser processing, due to the high cooling rates achieved.

Cladding with longer interaction times (0.3 and 0.45 s) and higher specific energies (58 and 87 MJ/m²) allows for extensive reactions between SiC and molten Al, due to the longer contact time between the two phases and a higher melt pool temperature. The fact that the only aluminium carbide observed in the solidification microstructure is Al₄SiC₄ shows that only reaction 2 occurred (see Introduction) and suggests that the melt pool temperature exceeded 1670 K during laser processing [6]. Al₄SiC₄ initially forms as faceted platelets at the interface between SiC and the liquid, whereas Si dissolves in the liquid [6]. Due to the morphology of Al₄SiC₄ crystals, this layer is not continuous and SiC remains in contact with the melt, thus allowing for the dissolution of a significant proportion of the SiC existing in the melt pool during the short lifetime of the liquid (0.3 and 0.45 s). Previous work carried out by one of the authors [25] on the kinetics of dissolution of a high melting temperature metal in molten Al showed that if a continuous layer of a solid product forms at the surface of the solid the dissolution time increases considerably. This is not the case here. According to Carotenuto et al. [26], the transport of Si and C in liquid Al is fast and the kinetics of the reaction $\text{SiC}_{(\text{s})} \rightarrow [\text{Si}] + [\text{C}]$ is controlled by the interface kinetics, which is relatively fast. The dissolution of SiC in molten aluminium leads to its enrichment in C and Si. As a result, Al₄SiC₄ precipitates directly from this liquid through a quasibinary peritectic reaction, as proposed by Oden and McCune [12]. This explains the presence of isolated Al₄SiC₄ platelets away from SiC particles (Figs. 3 and 6). The final concentrations of Si and C in the liquid depend on the amount of SiC dissolved, which increases with the contact time between the two phases and with temperature. Since contact time increases with interaction time and temperature increases with specific energy, the fraction of SiC dissolved will increase with these parameters. In fact, the proportion of SiC present in the clad microstructure decreases with increasing interaction time while the proportions of Al₄SiC₄ and Si increase.

After the precipitation of Al₄SiC₄, the Si-rich liquid presents a hypereutectic composition. Solidification proceeds by the precipitation of pro-eutectic silicon, initiated by heterogeneous nucleation on SiC and Al₄SiC₄ particles. In the case of coatings prepared with an interaction time of 0.3 s, as Si precipitates, the

composition of the liquid evolves towards the eutectic composition and the liquid eventually solidifies as α -Al+Si eutectic (Fig. 3d). In materials prepared with a longer interaction time (0.45 s) the Si particles grow to a significantly larger size and a Si-depleted zone forms around them, generating sufficient undercooling to create an α -Al halo with dendritic morphology around Si particles [23]. Eventually, the remaining liquid solidifies as α -Al+Si eutectic (Fig. 6c). Several authors have reported a similar solidification mechanism in laser cladding of hypereutectic Al-Si alloys [22,23,27].

The hardness of the coating containing 28 vol.% of SiC is 120 HV, in good agreement with the value reported by Sahin and Acilar [28] for an Al-10 wt.% Si/30 vol.% SiC bulk composite (130 HV). Despite their lower content of SiC particles (18 vol.%), coatings prepared using a higher specific energy present higher hardness (250 HV), due to the presence of considerable proportions of Al₄SiC₄ and Si (both phases having a hardness of about 1200 HV [11,22]).

The coatings present a better abrasive wear resistance than the substrate alloy due to the protection offered by the hard particles in their microstructure. The SiC particles of the abrasive medium easily indent and plough the Al-7Si alloy surface (90 HV), forming continuous parallel grooves. Material removal occurs by a microploughing mechanism, characterized by intense and repeated plastic deformation of the surface and the formation of ridges at the borders of the grooves [29]. The abrasive particles also groove the matrix of the composites, as shown in Fig. 10, but the morphology of the worn surface suggests that in this case plastic deformation is less intense than in the substrate alloy (Fig. 9), a conclusion consistent with the higher hardness of the material. It also suggests a larger contribution of the microcutting mechanism to the removal of material, indicated by the less prominent ridges around the grooves. This shift in wear mechanism may be explained by the smaller E/H ratio (E is the elastic modulus of the material and H is its hardness) of the harder composite matrix compared to the substrate alloy [30].

The coatings prepared with 330 MW/m² and 0.08 s (specific energy of 26 MJ/m²) present significantly higher wear resistance than those prepared with 193 MW/m² and 0.3 s (specific energy 58 MJ/m²), despite their lower hardness. According to Rabinowicz's equation [31] for abrasive wear, the wear rate is inversely proportional to the hardness of the material. Wear mechanisms in dual phase composite materials were discussed by Colaço and Vilar [32]. If the dispersed particles are much smaller than the abrasive particles and/or if the hardness of the reinforcement particles is lower than that of the matrix, the composite will behave as a homogeneous material and the material loss will be inversely proportional to its average hardness, provided that there is no change in the wear mechanisms. In this situation the hardness is given approximately by Khrushchov's linear law of mixtures [33] and the wear rate is inversely proportional to the volume fraction of reinforcement particles. However, a transition between microploughing and microcutting may occur when the E/H ratio decreases ([34], page 148), leading to an increase in wear rate. For the abrasive to be able to groove through the reinforcement particles the ratio

between their respective hardnesses (H_a and H_r) must be $H_a/H_r > 1.2$ ([34], page 136). In the present case, $H_a = H_r$, so the grooving action of the abrasive particles is interrupted by the reinforcement particles, decreasing the wear rate. Preferential removal of the matrix results in the reinforcement particles protruding from the surface after some sliding distance. These particles bear most of the load, leading to higher wear resistance [35,36].

The coatings prepared with high specific energy contain larger proportions of Al_4SiC_4 and Si precipitates and higher hardness than those prepared with lower specific energy. This is because these precipitates strengthen the material by the Orowan mechanism [37], so the hardening effect is inversely proportional to the average distance between precipitate particles. Therefore, coatings presenting large proportions of small precipitates have higher hardness. However, the hardness of Si and Al_4SiC_4 (1200 HV) is considerably lower than that of the abrasive, so they do not significantly enhance wear resistance. By contrast, although the SiC reinforcement particles do not significantly harden the material because the average distance between them is too large, they resist scratching by the abrasive, effectively contributing to the wear resistance of the coating.

5. Conclusions

The microstructure and properties of Al-12 wt.% Si/30 wt.% SiC coatings prepared by laser cladding depends strongly on the processing parameters, particularly power density and interaction time. For an interaction time of 0.08 s and a power density of 330 MW/m², corresponding to a specific energy of 26 MJ/m², SiC remains essentially undissolved. The microstructure of the coatings consists of SiC particles dispersed in a matrix of primary α -Al dendrites and interdendritic α -Al + Si eutectic. For higher specific energies (58 and 87 MJ/m²) SiC reacts and dissolves in molten Al, leading to a microstructure consisting of a small proportion of SiC, Al_4SiC_4 and Si particles dispersed in a matrix of α -Al + Si eutectic. The SiC particles concentrate at the bottom of the clad, where the temperature is lower.

The coatings prepared using specific energies of 58 and 26 MJ/m² present average hardness values of 250 and 120 HV, respectively. Despite its higher hardness, the wear rate of the former coating (1.7×10^{-4} mm³/m) is significantly higher than that of the latter (0.43×10^{-4} mm³/m).

Unlike Al_4SiC_4 and Si, SiC particles do not significantly harden the material, but they considerably improve its abrasive wear resistance. As a result, the processing parameters should be selected so that dissolution of SiC is avoided as much as possible.

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